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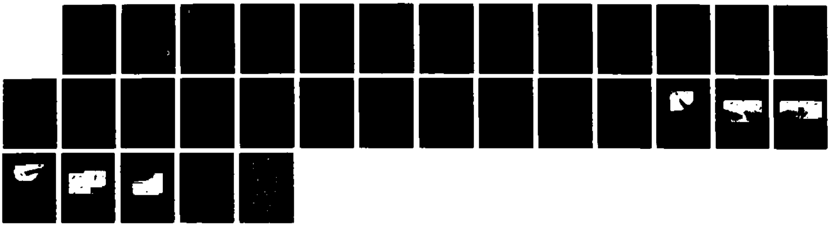

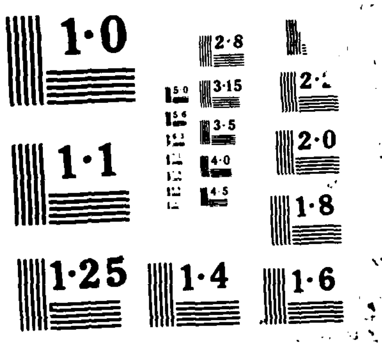
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on

Fundamental Studies of Beta Phase Decomposition Modes
in Titanium Alloys

by

H. I. Aaronson, Principal Investigator
A. M. Dalley, Graduate Student*
T. Furuhara, Graduate Student
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Fundamental Studies of Beta Phase Decomposition Modes in Titanium Alloys

H. I. Aaronson, A. M. Dalley, T. Furuhara and Y. Mou

ABSTRACT

An investigation of the interphase boundary structure of grain boundary alpha allotriomorphs in a Ti-7.15 W/O Cr alloy now in progress has established that complex, multiple sets of ledges are present at the interphase boundaries of allotriomorphs between not only the β grain with respect to which the allotriomorph is Burgers oriented but also with respect to the adjacent β grain, with which the allotriomorph necessarily has a non-Burgers orientation relationship. Questions are arising, however, as to whether certain systems of ledges are (immobile) structural ledges rather than (mobile) growth ledges. A tentative mechanism for the thickening of alpha allotriomorphs has been devised based upon interaction between structural and growth ledges. Resolution of the misfit dislocations which must be present on the terraces of both types of ledges and distinguishing between misfit dislocations and ledges with short risers are important problems which have arisen in TEM studies of these interphase boundaries. These problems are now being examined in detail. A comprehensive plan has been developed for the investigation of the massive transformation, (bcc)β → (hcp)ζ_m in a Ag-26 A/O Al alloy. This transformation, which is serving as a "stand in" for the crystallographically equivalent massive transformation previously discovered in several Ti-X systems with the support of this grant, is being utilized to make the first investigation of the interphase boundary structure associated with a massive transformation in any alloy system. Unlike Ti-X alloys, Ag-26 A/O Al permits retention of the matrix phase upon quenching to room temperature. The growth kinetics of the transformation in this alloy will be correlated with the interphase boundary structure in the context of the ledge mechanism. During this report period, the alloy was prepared and fabricated (by ALCOA) and the electropolishing procedure required for TEM studies was established.

1. INTRODUCTION

This program is an interlocking series of fundamental studies on the crystallography, morphology and kinetics of the proeutectoid alpha, the massive alpha and the bainite reactions in Ti-X alloys. Because that portion of the beta matrix which is not transformed to massive alpha is converted to martensite during quenching to room temperature in the Ti-X systems in which the $\beta \rightarrow \alpha_m$ reaction has been observed (1), however, the crystallographically equivalent massive transformation in a Ag-26 A/O Al alloy has been (2) and is currently being employed as a surrograte model massive transformation.

At least for the purposes of this program, studies of plate-shaped proeutectoid alpha crystals have been completed and the crystallography and interfacial structure of grain boundary allotriomorphs of proeutectoid alpha are now being intensively studied in a Ti-7.15 W/O Cr alloy as the Ph.D. thesis research of Mr. Tadashi Furuhara. The results of this work will be utilized to examine further the growth kinetics measurements of alpha allotriomorphs previously made during the Ph.D. thesis research of E. S. K. Menon (3). A counterpart study--also on bcc:hcp interfaces--is to be undertaken by Mr. Yiwen Mou on the Ag-26 A/O Al alloy as his Ph.D. thesis research. A similar correlation between interfacial structure and growth kinetics is to be made, in part by utilizing kinetic data previously reported on an alloy of nearly the same composition by Perepezko and Massalski (4).

During this report year, Miss Amber Daley completed her M.S. research on the interfacial structure of grain boundary alpha allotriomorphs and departed for a position with Energy Technology Consultants in Monroeville, PA. Mr. Furuhara has pressed on with this research and their first paper, reporting initial results from this program, is now being readied for submission to Scripta Metallurgica. Progress on the massive transformation studies has been considerably slower because Mr. Mou has been intensively engaged in completing most of his remaining coursework requirements and in preparing for his Ph.D. Qualifying Examination. He will be

taking this examination immediately after this report is submitted. Upon completion of the examination, he will devote most of his efforts to his research program. He will be making a substantial effort to complete his Ph.D. thesis by the time the present grant expires. It is hoped that Mr. Furuhara's work on the interfacial structure of grain boundary alpha allotriomorphs will be completed at the same time.

During the latter portion of Mr. Furuhara's research, this program will re-enter bainite studies as he examines the precipitation of $TiCr_2$ crystals at the interphase boundaries of proeutectoid alpha allotriomorphs in an effort to understand the crystallography, and by implication the interfacial energetics of the catalysis of $TiCr_2$ nucleation by the alpha allotriomorphs. Meanwhile, this grant is helping to support the P.I.'s accelerating paperwriting activities on the bainite reaction in steel. This ferrous-oriented effort is being actively aided by the results obtained on the bainite reaction in several Ti-X systems during the Ph.D. thesis research of Dr. Hwack Joo Lee and by the generalized understanding of the bainite reaction achieved during his studies. Four papers resulting from Dr. Lee's work will shortly appear in Acta Metallurgica and a fifth will be published in the Journal of Materials Science; page proofs for all five of these papers have been returned to the publishers. Our position in a major international symposium on the bainite reaction, to be held during the ASM International's World Metallurgical Congress in Chicago during the Fall of 1988, under the leadership of Prof. Morris Cohen of M.I.T., will be considerably strengthened as a result of the contributions to understanding the fundamental mechanisms of the bainite reaction made during Dr. Lee's thesis research.

2. CRYSTALLOGRAPHY, INTERFACIAL STRUCTURE AND GROWTH KINETICS OF GRAIN BOUNDARY ALPHA ALLOTRIOMORPHS IN A Ti-7.15 W/O Cr ALLOY

2.1 Introduction

Studies of the nucleation kinetics of grain boundary allotriomorphs of proeutectoid ferrite during the proeutectoid ferrite reaction in Fe-C (5) and in Fe-C-

X (6) alloys and during the proeutectoid alpha reaction in Ti-X alloys (3) have provided indirect but nonetheless rather convincing evidence that grain boundary allotriomorphs ought to be as coherent as possible during their nucleation stage. On the considerations of van der Merwe (7), during the early stages of growth most coherent interfaces--unless the misfit across them is very small indeed and/or mechanisms for acquisition of misfit dislocations are scarce or operating very slowly--should become partially coherent (8). This view contravenes earlier wisdom on grain boundary nucleated precipitates (9,10), according to which the allotriomorph:matrix interface is likely to be partially coherent with respect to only one of the matrix grains forming a grain face. A disordered structure is likely to form the interface between the allotriomorph and the adjacent matrix grain, since the orientation relationship between the allotriomorph and the latter grain was presumed to be irrational and thus incapable of supporting a partial coherent interfacial structure. However, measurements of the thickening kinetics of ferrite allotriomorphs in Fe-C alloys, wherein the collector/rejector plate mechanism (11) is unlikely to interfere, yielded rates somewhat less than those allowed by volume diffusion control (12), suggesting that partial coherency might reign on both broad faces of grain boundary allotriomorphs. Although experimental observations indicate that ferrite allotriomorphs have a rational orientation relationship with respect to only one austenite grain (13), facets are readily observed on both the interface of the allotriomorph with this austenite grain as well as the interface with the austenite grain to which the allotriomorphs are irrationally oriented. A similar observation was later made on grain boundary allotriomorphs of ξ_m formed during the $\beta \rightarrow \xi_m$ massive transformation in a Ag-26 A/O Al alloy (2). Hence the available indirect evidence on both nucleation and growth suggests that allotriomorphs are probably partially coherent with respect to both matrix grains forming grain faces. At the present time, however, no direct evidence is available in the literature on interfacial structure of grain boundary allotriomorphs. The present investigation was undertaken to fill this important gap in our knowledge and will serve as the Ph.D. thesis of Mr. Tadashi Furuhara.

Although it might have been preferable to have conducted this investigation on the proeutectoid ferrite reaction, since this is by far the most widely known and studied one in solid metallic alloys, conversion of effectively all of the austenite untransformed during isothermal reaction to martensite during quenching to room temperature make observation of the interphase boundary structure of ferrite allotriomorphs in hypoeutectoid steels impossible with room temperature techniques. (Use of hot-stage TEM could circumvent this problem, but the loss of resolution attending observations made at elevated temperatures would have even more serious consequences for ferrite allotriomorphs than for alpha allotriomorphs because of the larger misfits between the bcc and fcc than between the bcc and hcp lattices. It will shortly be seen that even with alpha allotriomorphs resolution of all of the interfacial structure present is posing serious problems.) On the other hand, in numerous hypoeutectoid Ti-X alloy systems it is possible to retain all of the beta matrix upon quenching to room temperature, even in thin foils, once a critical composition is exceeded. We chose a hypoeutectoid Ti-7.15 W/O Cr alloy for the present study because we have used the same alloy in several previous investigations and have found it quite satisfactory for the purposes of the present study. Some very small precipitates of the transitional omega phase do form during quenching to room temperature, but they do not seriously interfere with TEM observations.

2.2. Experimental Procedures

This alloy was obtained from Titanium Metals Corporation of America, Henderson, NV. It was homogenized for 3 days at 1000°C. The homogenized alloy was then hot rolled by TIMET and the product was given the same homogenization treatment. It was immediately recognized that the coarse beta grain size characteristic of this alloy (and most other Ti-X alloys capable of being converted 100% to the beta phase) would make a TEM study of grain boundary alpha allotriomorphs infeasible. Dr. Donald Kroger of the Oak Ridge National Laboratory solved this problem for us by rapidly solidifying small portions of the homogenized

alloy by means of the hammer and anvil technique. This yielded a beta grain size a few tens of microns in diameter and ensured that most thin areas, prepared for TEM, would contain portions of at least a few beta grain boundaries. Specimens cut from the rapidly quenched material were wrapped in Ta foil and encapsulated in Vycor under a vacuum $<10^{-5}$ torr. They were transformed by upquenching directly to the intended isothermal reaction temperature. During this process, the omega phase precipitates must have dissolved prior to allotriomorph nucleation and growth, since it is a long familiar observation that in the presence of omega the alpha phase nucleates preferentially at omega:beta interfaces, yielding an ultra-fine dispersion of minute alpha crystals. To date, nearly all studies have been performed at a reaction temperature of 700°C. TEM specimens were prepared by means of ion milling, since electropolishing has been convincingly shown to produce the "interface phase" (14), which grows allotriomorphically along alpha:beta boundaries and hence would entirely prevent this investigation from being conducted. The orientation relationships between alpha allotriomorphs and their bounding beta grains were determined by analyzing the Kikuchi pattern obtained from each phase. At least two different tilting conditions were used for a given phase. The error associated with this procedure was less than 1°.

2.3. Results

Grain boundary allotriomorphs were found to have an orientation relationship which is either exactly or very close to the Burgers orientation relationship with only one bounding beta matrix grain $(0001)_\alpha // (0\bar{1}1)_\beta$, $[11\bar{2}0]_\alpha // [111]_\beta$. The orientation relationships with respect to the adjacent beta grain was generally irrational.

Fig. 1a shows a typical grain boundary alpha allotriomorph (labelled α_1 in this Figure). In Fig. 1b, the major poles of both bounding beta grains are plotted on a standard hcp stereographic projection, i.e., in terms of the lattice of the alpha allotriomorph. α_1 is related to the left-hand beta grain, β_B, with the Burgers

orientation relationship, as indicated in Fig. 1b. However, the orientation relationship with respect to the right hand beta grain, β_{NB}, is seen to be nearly 20° away in terms of parallel directions and 3° away in terms of parallel planes from the Burgers relationship (and hence is called "non-Burgers (NB) oriented"); this is thus a fairly irrational orientation relationship.

The interphase boundary between α_1 and its β_B grain is shown in Figures 2a and b. Fig. 2a illustrates a set of high ledges, irregular in both spacing and path. Fig. 2b, taken of the same area but employing a different diffraction vector shows that the terraces of the ledges in Fig. 2a contain another set of much smaller ledges. These ledges are much more closely and uniformly spaced and lie in a well defined direction. The arrows in Figs. 2a and b point out the same area within these Figures; the same high ledge is visible in both, but the smaller ledges can be seen only in Fig. 2b. While some of the higher ledges are roughly orthogonal to the smaller ledges, other high ledges, Fig. 2b indicates, share the same habit plane as the smaller ledges. Because of their small and regular spacing and approximately linear paths, the possibility may be suggested that the smaller family of ledges may be of the structural (15) rather than the growth type, exisiting to improve misfit compensation across the interphase boundary but immobile with respect to growth.

Fig. 3 shows the interfacial structure between the same grain boundary allotriomorph and the beta grain with respect to which it has a non-Burgers orientation relationship. The same area is shown in dark-field (Fig. 3a) and in bright-field (Fig. 3b). Note that widely spaced ledges in Fig. 3a, whose height is estimated from an equation due to Gleiter (16) as 4 nm., are succeeded by parallel ledges which are much more closely spaced as a change occurs in the orientation of the alpha:beta boundary. Both Figs. 3a and b show that at least one other set of ledges, much more widely and irregularly spaced, is also present. In the vicinity of the single arrowhead in Fig. 3a, the closely spaced ledges are seen to be abruptly shifted when crossed by the more widely spaced ones. The kinks thus formed support the view

that the irregularly spaced features are indeed ledges. The region on the left-hand side of Fig. 3b provides clear indications, particularly from extinction contour displacements, that the widely spaced "vertical" features are also ledges; hence their parallel, much more closely spaced counterparts should also be ledges.

Fig. 4 shows another grain boundary allotriomorph, labelled α_2. Orientation relationships with respect to the two bounding beta grains are indicated on the stereographic triangle included in this Figure. The orientation relationship with respect to the lower beta grain is seen to be exactly of the Burgers type whereas that with respect to the upper beta grain in the micrograph is about 10° away from the Burgers relationship in both planes and directions. However, closely spaced ledges are visible around almost all of that portion of the perimeter of the allotriomorph which is contact with the beta matrix. With suitable tilting, ledge structures are disclosed at the remaining orientations of the interphase boundary. Fig. 5 is a dark-field view at higher magnification of a portion of the α_2:β_B interface shown in Fig. 4. Two sets of ledges, labelled A and B, non-uniformly spaced and irregularly aligned, are clearly shown in this micrograph. From application of the Gleiter equation, both sets of ledges were found to be approximately 7 nm. high. Closer observation of this interfacial structure discloses a third set of defects, marked C. Where these defects intersect the ledges of sets A and B, kinks are formed with the same spacing, c. 4 nm., as the inter-defect spacing of set C. A $\underline{g}\cdot\underline{b}$ analysis was conducted upon the defects of set C, with results as summarized in Table I. Assuming that the criterion for invisibility is solely $\underline{g}\cdot\underline{b} = 0$ and that the defects are perfect dislocations, these defects are seen to correspond to $\underline{b} = \underline{a}/2[\bar{1}\bar{1}\bar{1}]$. Although this vector could lie either within or at an angle to the (as yet unknown) terrace planes of ledge sets A and B, the aforementioned kinks suggest that defects are in fact ledges with very low risers.

The interface between α_2 and the beta grain with respect to which it is not Burgers related, β_{NB}, is shown in dark-field and at higher magnification in Fig. 6.

Two arrays of defects, whose direction and average spacing are indicated by appropriately oriented pairs of arrows, are displayed. Extinction contour displacements indicate that both sets are probably ledges. The spacing between parallel ledges in both set is c. 5nm.

2.4. Discussion

The principal microstructural feature requiring discussion is the one which did not appear, namely misfit dislocations. TEM investigations of interphase boundary structure in many alloy systems have shown that misfit dislocations are invariably present on the terraces of ledges (17,18). By their nature, the terraces of ledges must be regions in which matching between the crystal lattices is exceptionally good. However, only rarely is the lattice matching across terraces so exact that the terraces are fully coherent. Hence at least one and usually two arrays of misfit dislocations should appear on the terraces. Presumably the primary reason why misfit dislocations have not so far been observed with certainty in the present study is that they are obscured by the contrast between (often) closely spaced ledges. In order to detect (and quantitatively identify) the misfit dislocation structure on the terraces of ledges, it will be necessary to use the weak-beam, dark-field technique (19) and to employ diffraction vectors, probably in the alpha phase, which yield maximum visibility of misfit dislocations whose Burgers vector lies in the plane of the terraces.

Another problem of a fundamental nature which this study brings up is that of structural ledges. These were first predicted to develop at $\{111\}_{fcc}$ $\{110\}_{bcc}$ interfaces because the periodic insertion of a one-, two- or three-plane high ledge will cause regions of optimum matching to repeat, <u>laterally</u> <u>displaced</u>, on the terrace of the ledge thereby introduced (15,20). A consequence of the introduction of structural ledges is that the density of misfit dislocations required is reduced. The tri-atomic version of the predicted structural ledges have been found and unequivocally identified on the broad faces of ferrite plates in an Fe-C-Si alloy (21).

However, an equally careful study of facets on Cr-rich precipitates in a dilute Cu-Cr alloy yielded two sets of misfit dislocations without any structural ledges at all but perhaps one of the facets closely examined (22). A much less well documented claim that structural ledges are absent has also been made at fcc:bcc boundaries in a complex duplex stainless steel (23). Recently, Hackney and Shiflet (24) have reported structural ledges on facets upon θ precipitates in an Al-4% Cu alloy; identification as structural ledges was evidently based upon immobility at temperature when observed with hot-stage TEM. During an on-going study, structural ledges have also been suspected of being present at the broad faces of cementite plates in an Fe-C-Mn alloy (25). The evidence for the latter deduction is essentially circumstantial: closely and uniformly spaced and reasonably linear ledges. Because the precipitate lattices involved in both the Al-Cu and the Fe-C-Mn alloys are exceedingly complex, a crystallographic investigation of the structural ledge problem in them may not be particularly rewarding. In the present situation, however, the lattices of both phases are simple. Hence, we have begun to repeat the original studies of Hall et al (15) on bcc:fcc interfaces upon bcc:hcp interfaces, initially assuming the Burgers orientation relationship. Our initial objective will be to ascertain whether or not the areas of optimum matching will repeat when one-, two- and three-atom plane high ledges are introduced at the Burgers interface. We also plan to attempt some experiments designed to ascertain whether or not the ledges in question are mobile. Because of the extreme reactivity of Ti alloys, hot-stage TEM will not be employed. Instead, thin foils will be carefully re-encapsulated in internally gettered Vycor capsules and then held for brief periods of time at the same reaction temperature as that initially employed. Provided that areas previously photographed can again be located, even small amounts of motion should be readily detectable. Particularly since numerous ledges which are obviously of the growth type (characterized on the basis of their resemblance to growth ledges in other alloy systems) are present, the motion of such ledges will serve to calibrate the mobility or immobility of the sets of ledges suspected of being of the structural type.

On the basis of observations so far made upon the ledge structure on the broad faces of alpha allotriomorphs, the schematic drawing of such a structure, shown as Fig. 7, was constructed. The bases for this model are: at least two sets of growth ledges are present; and, since the risers of growth ledges are often parallel to those of (suspected) structural ledges, these risers are immobile. Hence migration of the risers of growth ledges must be accomplished by means of kinks on the risers. Intersection of systems of ledges is one possible source of kinks; others may also be operative. Note that this model implies that there will be minimal interference between the diffusion fields of adjacent ledges in the same or in different systems of ledges because atomic attachment sites are confined to small-area kinks.

2.5. Future Plans

For sometime to come, TEM will be emphasized, with efforts increasingly focused upon detecting and quantitatively characterizing misfit dislocation arrays on the terraces of growth ledges. Efforts will also be made to describe more quantitatively the growth ledges now being observed. Modeling studies are also being initiated at this time, however, in order to examine the concept of structural ledges in the context of bcc:hcp interfaces, initially assuming that the Burgers orientation relationship is operative.

Once we begin to observe repeatedly the misfit dislocation structures which ought to be present on the terraces of ledges, we plan to perform O-lattice modeling of the dislocation structure of bcc:hcp interphase boundaries. We plan to conduct such studies at a wide range of boundary orientations, assuming the Burgers lattice orientation relationships, and to attempt correlations between the misfit dislocation structures observed and those predicted. In order to make these comparisons, however, it will be necessary not only to detect and characterize the misfit dislocation sturctures but also to ascertain the conjugate habit planes of the terraces at which they appear.

In later studies, the O-lattice modeling will be conducted on the basis of experimentally observed non-Burgers orientation relationships. Non-Burgers interfaces which we are able to characterize accurately will then be analyzed on this basis.

At the present time, there is very little predictive capability available, for any pair of crystal structures, on the formation frequency of growth ledges (though not of structural ledges). However, since the present investigation will permit examination of an unusually large number of boundary orientations, we shall "keep an eye open" for possibilities of correlating inter-ledge spacing with boundary orientation.

3. MASSIVE TRANSFORMATION IN A Ag-26 A/O Al ALLOY

3.1 Introduction; Research Plan

This investigation will comprise the Ph.D. thesis of Mr. Yiwen Mou, who is a faculty member on leave from Chongqing University in the PRC. This study will presumably be the first in which our view (26) that the interphase boundaries of crystals produced during a massive transformation are frequently partially coherent, rather than disordered as originally predicted by Massalski (27). The alloy composition we are utilizing for this investigation, Ag-26 A/O Al, has been previously studied by Plichta and Aaronson (2), who succeeded only in ascertaining orientation relationships of ζ_m crystals with respect to their matrix β grains, using the selected area electron channeling technique. Among the 47 ζ_m crystals examined in this manner (all of which nucleated at β grain faces or edges), 46 were found to have a Burgers orientation relationships with respect to at least one of the β grains with which they were in contact. Irrational orientation relationships were again found in many instances with respect to the other β grain(s), but the observation that facets appeared even on the non-Burgers interfaces strongly suggested to us that these interfaces were also likely to be partially coherent. Our attempts to observe interphase boundary structure were frustrated, however, by an unfavorable texture of

the foils prepared (which caused too many interphase boundaries examined to lie nearly parallel to the electron beam, where their structure could not be observed) and, above all, by the precipitation, at room temperature, of the μ phase, evidently by means of another massive transformation. We plan to circumvent this problem in the present investigation by the use of a cold stage attachment available for our Philips 420 TEM.

Since it is our belief that the massive transformation is as governed by the considerations of Gibbs and successors, particularly with regard to the role of crystallography in nucleation kinetics, as is precipitation from solid solution, we are anticipating that the interphase boundary structures observed on ξ_m crystals will be closely similar to those which Mr. Furuhara has been finding on the broad faces of grain boundary alpha allotriomorphs in Ti-7.15% Cr. Inasmuch as essentially all of the ξ_m crystals of interest during this investigation will be grain boundary nucleated, the research plan developed for the Ti-Cr alloy will also be utilized for the $\beta \rightarrow \xi_m$ transformation in Ag-26 A/O Al.

3.2 Alloy Preparation

An ingot of Ag-24.4 A/O Al was prepared for us by the ALCOA Research Laboratories in ALCOA Center, PA. However, this alloy proved to be too brittle to roll, quite possibly because of rapid precipitation of the μ phase. The ingot was then remelted as a Ag-26 A/O Al alloy, comparable to the one previously employed. Little difficulty was now encountered in rolling out sheets of alloy 0.3 mm. thick. We would have preferred the 24.4 A/O Al composition because this one lies exactly at the congruent composition in this system and thus there could be no complaint about possible transformation with the fcc + hcp region; however, this proved impracticable--and we do not think that such complaints will be serious.

3.3 Experimental Procedures

All necessary heat treatment procedures for this study have been previously reported (2). It remained to develop a suitable procedure for the preparation of thin foil specimens for TEM. An ethanol solution containing from 5 to 15% perchloric acid was found to be satisfactory. In order to avoid the martensitic transformation--which occurs at temperature below that of the μ phase transformation--temperatures in the vicinity of 0°C were employed. Good results were obtained at a voltage of 15 - 25 V and a current of 8 - 20 mA.

Mr. Mou is now studying for his Ph.D. Qualifying Examination. When this obstacle has been surmounted, he will be ready to begin immediately the heat treatment of his specimens and their TEM examination to ascertain the interphase boundary structures operative.

PUBLICATIONS, HONORS AND AWARDS

During this report year, the following papers, based upon research supported entirely or in part by our AFOSR grant, have been published or accepted for publication (and are thus "in press").

1. E.S.K. Menon and H.I. Aaronson, "Morphology, Crystallography and Kinetics of Sympathetic Nucleation", Acta Met., 35, 549 (1987).

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At the Fall 1987 ASM Awards Banquet, held during ASM's Fall Meeting in Cincinnati, OH, the P.I. was presented with the Albert Sauveur Achievement Award. Research on the bainite reaction was emphasized in the citation for this award.

The P.I. presented an invited paper on the fundamentals of diffusional nucleation and growth at a Workshop on "Competing Interactions and Microstructures: Statics and Dynamics", held at the Los Alamos National Laboratory in May, 1987, a keynote paper on "The Bainite Reaction" at the International Phase Transformations Conference held at Cambridge University in July, 1987, the opening paper at a Phase Transformations Conference held in conjunction with an Analytical Electron Microscopy Meeting which took place immediately afterwards in Kona, Hawaii (this paper attempted to cover the entire field of diffusional phase transformations within the compass of a 30 min. presentation!), an invited paper on the proeutectoid ferrite reaction in Fe-C and Fe-C-X alloys at a symposium in honor of Prof. J. S. Kirkaldy at McMaster University and an after-dinner address at the Cahn Materials Science Symposium held at the National Bureau of Standards, Gaithersburg, MD in January, 1988, in honor of Dr. Cahn's 60th birthday. AFOSR support was acknowledged in all of these presentations and papers. In addition, the P.I. was the technical speaker at meetings of the Detroit Chapter of TMS and the Pittsburgh Chapter of TMS; both

talks dealt with the proeutectoid alpha and the eutectoid reactions (mainly bainite) in Ti-X alloys.

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Table 1: Results of $\underline{g} \cdot \underline{b}$ Experiments on Defect Set C in $\alpha_2:\beta_B$ Interface (Fig. 5)

$\vec{g}_{bcc}$	Observed Contrast	$\vec{g} \cdot \vec{b}$ for $\frac{a}{2}[1\bar{1}1]$
$0\bar{1}1$	*strong*	2
101	*weak*	2
110	*none*	0
$\bar{1}12$	*strong*	-4

FIGURE CAPTIONS

Fig. 1. An allotriomorph (α_1) formed at a β grain face in Ti-7.15 w/o Cr, reacted for 20 min. at 700°C.

a) Bright-field TEM micrograph.

b) (0001) hcp stereographic projection of the allotriomorph α_1 showing orientations of the adjacent β grains relative to $(0001)_{\alpha_1}$ β_B--Burgers related β grain. β_{NB}--non-Burgers related β grain.

Fig. 2. TEM micrographs showing the interfacial structure of the allotriomorph α_1 with the β_B grain.

a) dark-field $\vec{g} = [\bar{1}101]_{\alpha_1}$;

b) weak-beam, dark-field $\vec{g} = [0002]_{\alpha_1}$.

Fig. 3. TEM micrographs showing the interfacial structure of the allotriomorph α_1 with the β_{NB} grain.

a) weak-beam dark-field $\vec{g} = [\bar{1}10\bar{1}]_{\alpha_1}$;

b) bright-field in the two beam condition: $\vec{g} = [000\bar{2}]_{\alpha_1}$

Fig. 4. An example of allotriomorphs (α_2) formed at another β grain boundary in Ti-7.15 w/o Cr reacted for 20 min at 700°C.

a) Bright-field TEM micrograph,

b) (0001) hcp stereographic projection of the allotriomorph α_2 showing orientations of adjacent β grains relative to $(0001)_{\alpha_2}$.

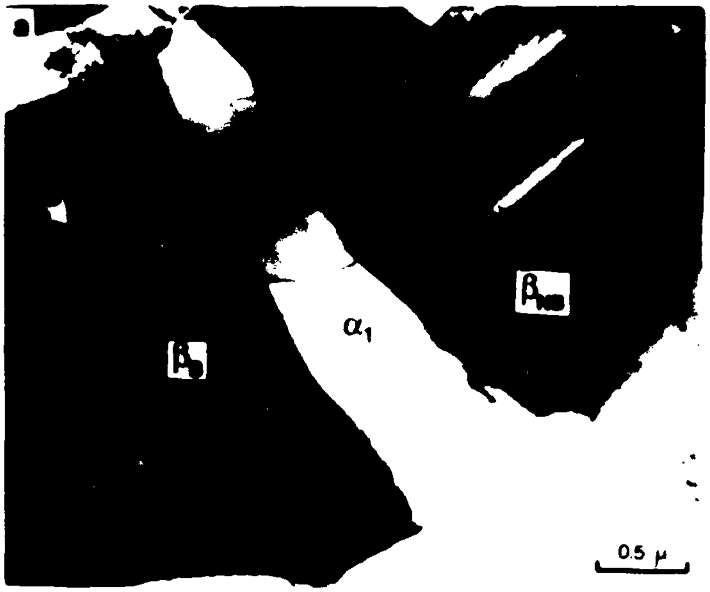
β_B -- Burgers-related β grain

β_{NB} -- non-Burgers-related β grain

Fig. 5. Dark-field micrograph showing the interfacial structure of allotriomorph α_2 with the β_B grain ($\vec{g} = [0\bar{1}1]_{\beta_B}$), and a schematic drawing indicating orientations of the defects at the interface.

Fig. 6. Bright-field micrograph showing the interfacial structure of the allotriomorph α_2 with the β_{NB} grain, in the two-beam condition: $\vec{g} = [10\bar{1}1]_{\alpha_2}$.

Fig. 7. Schematic diagram illustrating a growth ledge/structural ledge configuration for two sets of non-parallel growth ledges.


a
α_1
β_{NB}
β_B
0.5 μ

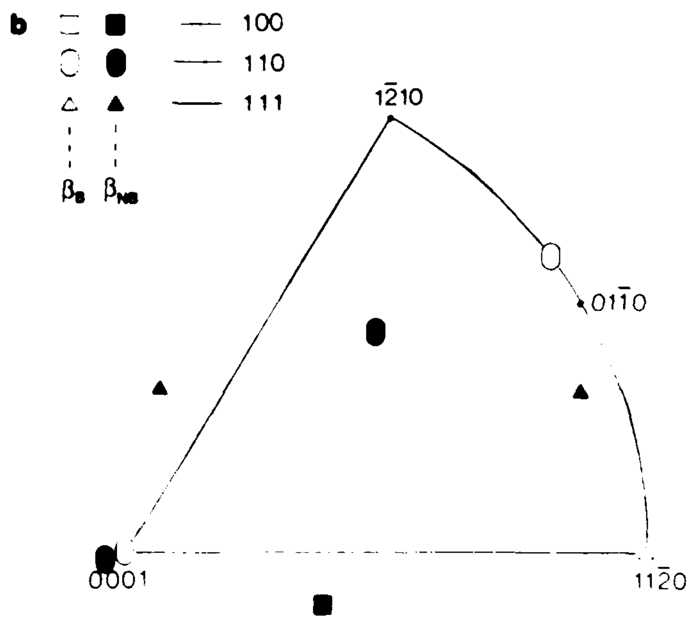

b
100
110
111
β_B
β_{NB}
$1\bar{2}10$
$01\bar{1}0$
0001
$11\bar{2}0$

Figure 1

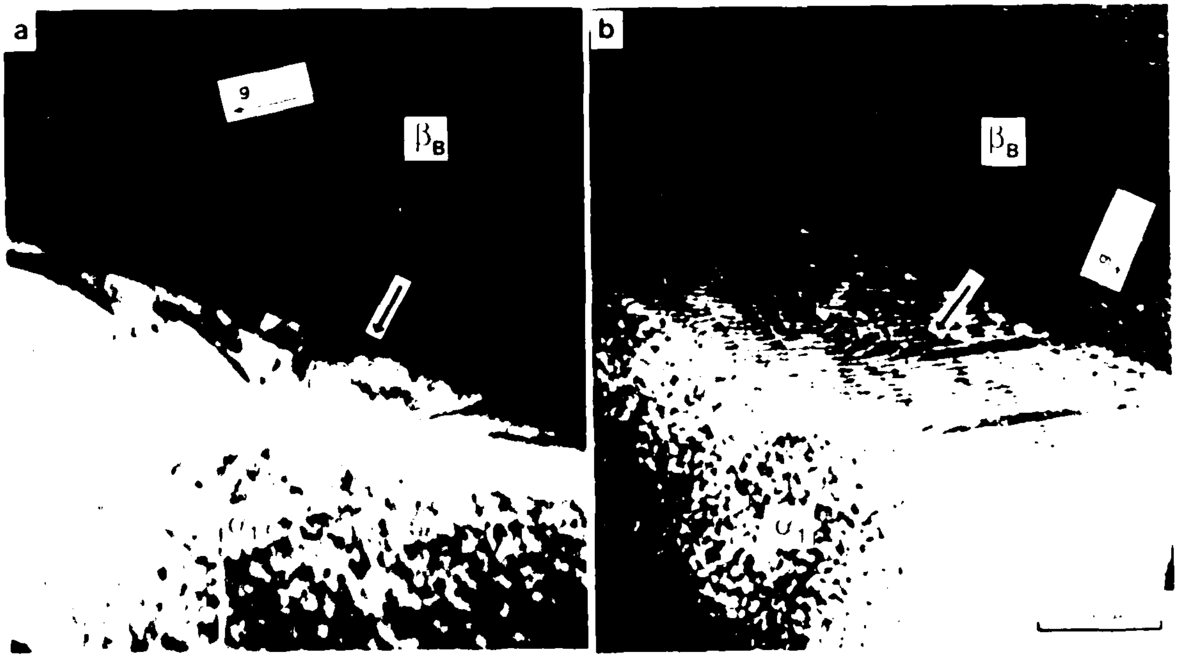
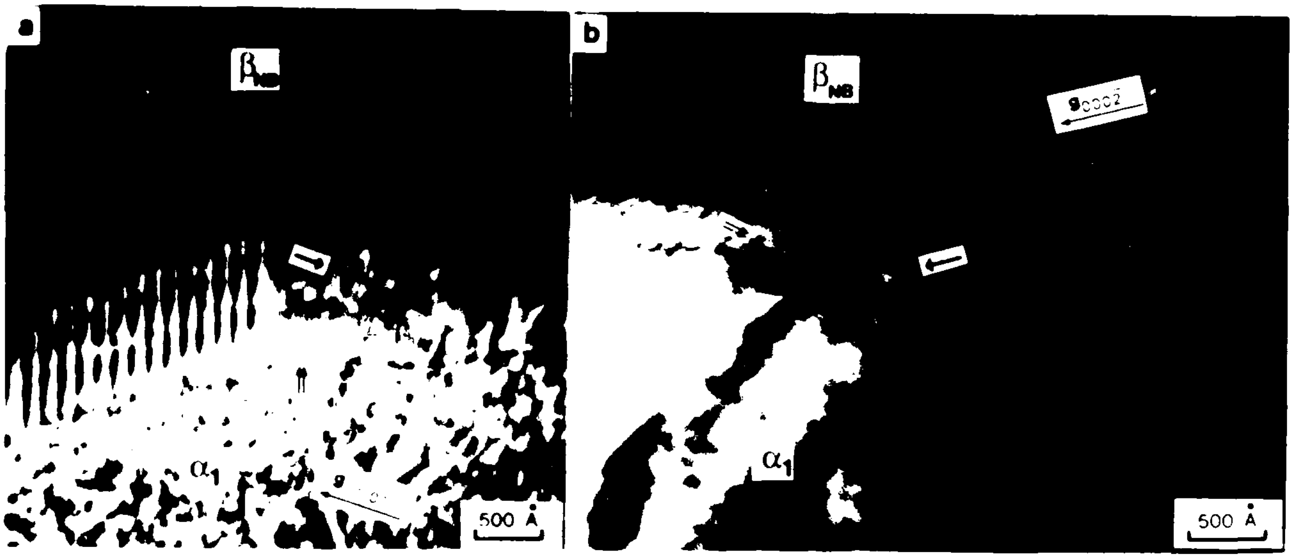


Figure 2



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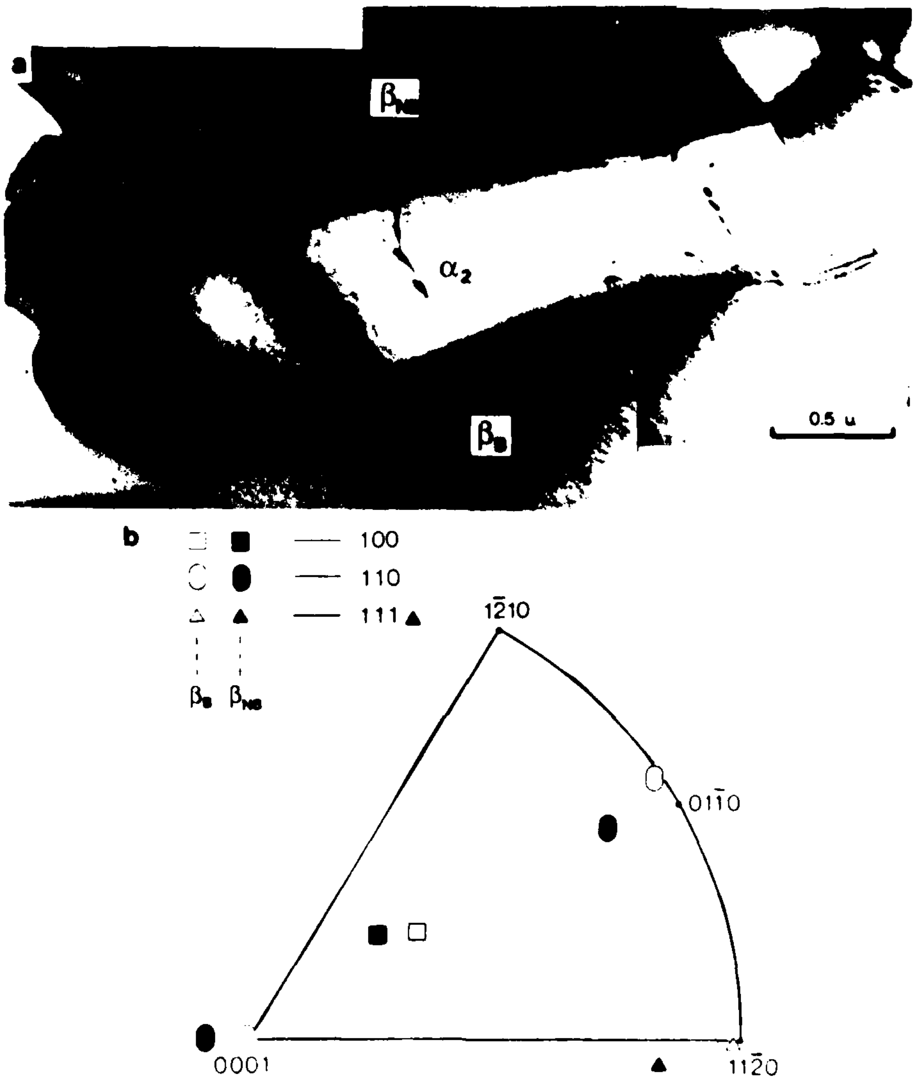

a
β_{NS}
α_2
0.5 u
β_S
b
100
110
111
β_S β_{NS}
$1\bar{2}10$
$01\bar{1}0$
0001
$11\bar{2}0$

Figure 4

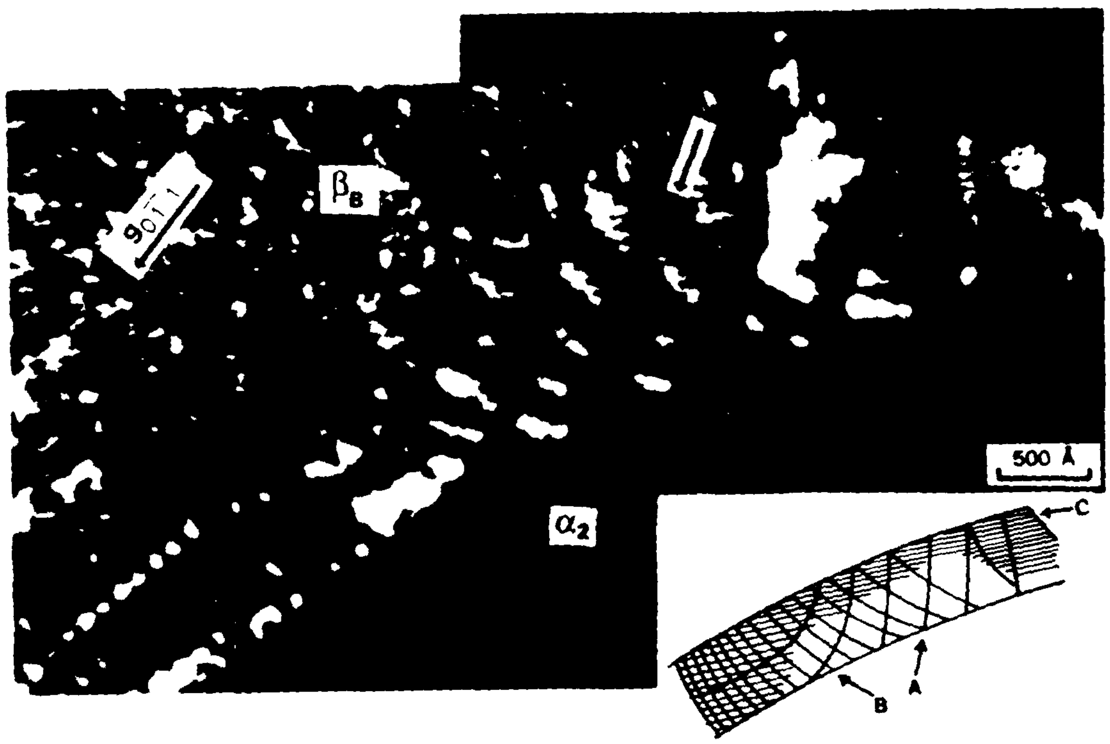

Figure 5

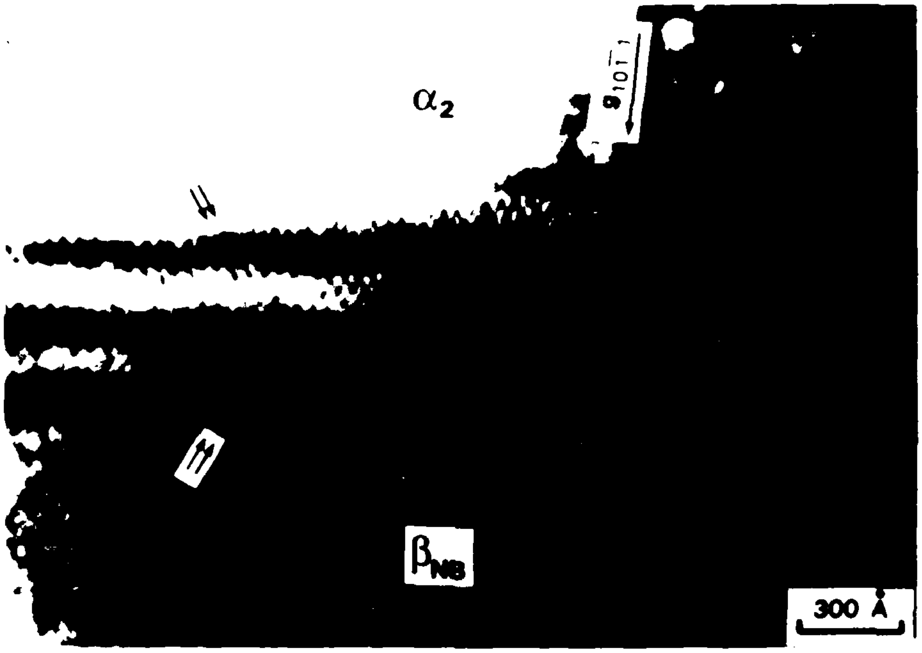

Figure 6

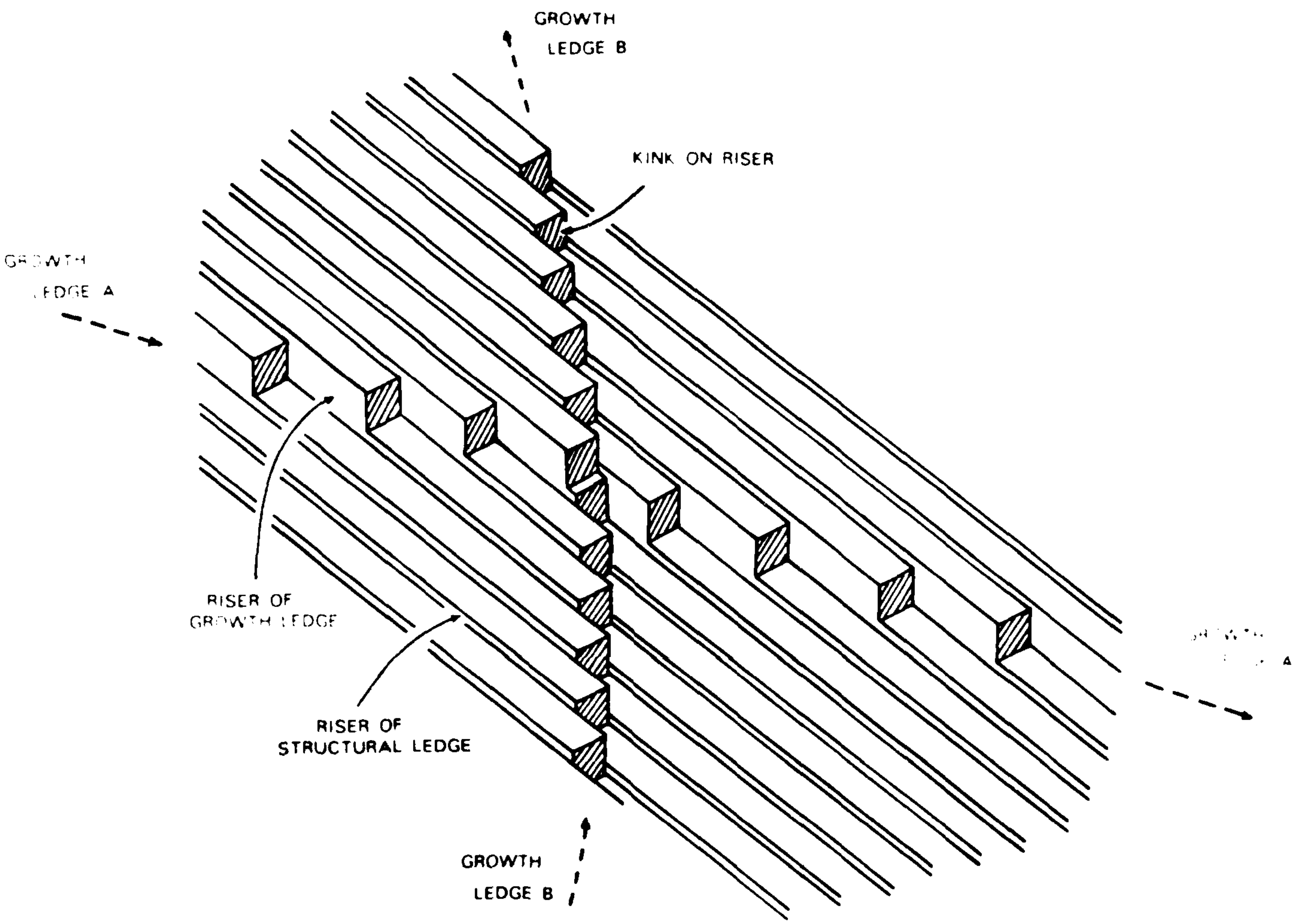
GROWTH
LEDGE B
KINK ON RISER
RISER OF
GROWTH LEDGE
RISER OF
STRUCTURAL LEDGE
GROWTH
LEDGE B

Figure 7

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